

# Optimal temperature profiles for minimum residual stress in the cure process of polymer composites

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## Abstract

Manufacturing of polymer composites using a curing process requires the specification of the temperature as a function of time, i.e., the temperature profile. It is of utmost importance that the selected profile satisfies a number of criteria which include the minimum residual stresses, minimum cure cycle time and full degree of cure. The development of residual stresses during the cure cycle is one of the most important problems as they affect the strength and the mechanical properties of the final product adversely. The object of the present study is to determine the optimal temperature profiles used during curing in order to minimise these stresses. Numerical simulation is used to study the development of stresses during curing based on a process model which includes the effects of chemical and thermal strains and the viscoelastic material behaviour. The process model is implemented to conduct a parametric study to observe the trends and characteristics of the residual stress history varying engineer controllable input parameters. The gradients of the applied temperatures at different dwell times are identified as essential process parameters. An optimised curing cycle based on this observation is developed using the results of the parametric study. The optimal cycle achieves substantial reduction in the residual stresses and curing time for fully cured composites as compared to manufacturer recommended cycles. © 1999 Elsevier Science Ltd. All rights reserved.

*Keywords:* Optimal cure cycle; Residual stresses; Cure temperature profiles; Polymer composites; Curing processes

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## 1. Introduction

One of the important processes which are used to produce high-performance laminated composites out of thermosetting-matrix composites is the autoclave curing of the prepreg lay-ups. This process, in general, involves exposing a multi-layered fibre–resin mixture to elevated temperatures and pressures for prescribed periods of time. The high temperatures initiate and promote cross-linking of polymer while the applied pressure helps remove any gases and voids from the mixture and leads to the compaction of the specimen.

The constituent materials of the composite react differently to the temperature changes during the processing. Chemically, the reinforcing fibres are affected very little during the process cycle. However, the polymer matrix contracts during cross-linking by as much as 6%. Also the thermal deformation of fibres is quite small during the cooldown period. The polymer matrix on the

other hand has a higher coefficient of thermal expansion resulting in higher deformation. These deformations are balanced internally within the composite by the inducement of residual stresses. In fact residual stresses can be high enough to cause cracking within the matrix even before the mechanical loading is applied as noted by White and Hahn [1]. This microcracking adversely affects the strength of the material and exposes the fibres to chemical degradation [2]. Even though these stresses may not reach this level, strength is still adversely affected by the presence of residual stresses since the component has been preloaded during the curing process [2].

Consequently, the magnitude and the duration of the applied temperature during the curing cycle is a significant process parameter affecting the strength and integrity of the product since the extent of the thermally induced stresses depend on the temperature profile. However, cure cycles can be optimised by a judicious choice of the temperature profiles to produce a fully cured material with minimum residual stress.

Selection of cure cycles in practice is mostly based on a trial-and-error procedure where either numerical

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simulations or experimental trials are conducted on several candidate cycles. However, a trial-and-error approach is inefficient, expensive and does not even ensure the selection of the best temperature profile for a particular application. This is because there are many interrelated factors to be considered when choosing a cure cycle. For example processing at low temperatures would reduce thermally induced stresses, however, a minimum temperature has to be attained before cross-linking can start and the degree of the cure affects the mechanical properties of the material as shown by Lee and Springer [3]. Besides, the cure cycle takes longer at low temperatures adding to the cost of manufacturing. Higher temperatures may speed up the curing process but they will increase thermally induced residual stresses.

Several studies were directed to the numerical simulation of the curing process with a view towards elucidating various aspects of the cure process. A review of various physical models is given by Thomas et al. [4] who provide an extensive set of experimental results on cure cycle effects.

Recently, there have been a number of studies directed to the optimization of the cure cycle with the objective of reducing the process-induced residual stresses (see Refs. [5–8]). In these studies, experimental techniques and numerical simulations were used to optimise the cure cycle. In other optimization studies, different objectives were considered such as cure cycle times by Rai and Pitchumani [9] and consolidation for thick laminates by Chang et al. [10].

In the present study, the cure cycle optimization is based on the effect of the temperature gradient on the development of residual stresses. The numerical simulation results identify the gradient of the temperature profile as an effective process parameter which has been neglected in the previous optimization studies. In fact the parametric study gives valuable insight into the understanding of the stress responses to variations in temperature profile. The results of this study are used to develop an optimised cure cycle with reduced residual stresses which can be in the order of 30% as compared to recommended cure cycles. Moreover the optimal temperature profile also improves the cycle time.

## 2. Basic equations

The mathematical model of the reaction kinetics can be of phenomenological or mechanistic. In the present study a phenomenological model is adopted which describes the reaction process in sufficient detail without considering reaction of the individual species with each other as noted by White and Hahn [11]. In particular, an expression of the general form [11,12]

$$\frac{d\alpha}{dt} = Kf(\alpha), \quad (1)$$

will be used to model the reaction rate. In Eq. (1)  $\alpha$  is the degree of cure,  $t$  is the reaction time,  $K$  is a reaction rate constant, and  $f(\alpha)$  represents some function of  $\alpha$ .

Various specific forms of Eq. (1) can be used to model the cure kinetics of thermosetting resins. This form depends on the required level of sophistication needed to study a given problem with sufficient accuracy. For the present problem the model developed by Bogetti and Gillespie [13] to describe the response of unsaturated polyester resins will be used which is given by

$$\frac{d\alpha}{dt} = A \exp\left(\frac{-\Delta E}{RT}\right) (1 - \alpha)^n \alpha^m, \quad (2)$$

where  $A$  is the frequency factor,  $\Delta E$  the activation energy,  $R$  the universal gas constant,  $T$  the temperature and  $m$  and  $n$  are cure kinetics exponents. The cure kinetic parameters  $K$ ,  $m$ , and  $n$  can be determined by thermal analysis techniques such as Differential Scanning Calorimetry (DSC).

The mechanical properties of the composite material change as the curing progresses. In particular, the transverse compliance,  $S_{22}(t)$ , undergoes a substantial change with time during cure (see Weitsman [14]). This behaviour can be modeled by a power law of the form

$$\bar{S}_{22}(\alpha, t) = \bar{S}_{22i}(\alpha)f(t) + D(\alpha)\left(\frac{t}{a_T(\alpha, T)}\right)^{q(\alpha)}, \quad (3)$$

where  $\bar{S}_{22}$  is the transverse compliance,  $\bar{S}_{22i}$  the initial transverse compliance,  $f$  a material dependent function chosen to agree with the results of White and Hahn [11],  $D$  the transverse creep coefficient,  $a_T$  the shift factor and  $q$  is the transverse creep exponent. The creep parameters  $D$  and  $q$  can be obtained through transverse creep experiments. The initial compliance  $\bar{S}_{22i}$  corresponds to the elastic response of the material. The creep parameters  $D$ ,  $q$ , and  $a_T$  depend on the degree of cure and can be represented as linear functions of  $\alpha$ , viz.

$$D(\alpha) = D_i + (D_f - D_i)\alpha, \quad (4)$$

$$q(\alpha) = q_i + (q_f - q_i)\alpha, \quad (5)$$

where  $D_i$  is the uncured transverse creep coefficient,  $D_f$  the fully cured transverse creep coefficient,  $q_i$  the uncured transverse creep exponent and  $q_f$  is the fully cured transverse creep exponent [11].

The initial transverse modulus of elasticity  $\bar{E}_{22}$  can be expressed as

$$\bar{E}_{22i}(\alpha) = E^*, \quad 0 \leq \alpha < \alpha^*, \quad (6a)$$

$$\bar{E}_{22i}(\alpha) = a_0 + a_1\alpha + a_2\alpha^2, \quad \alpha^* \leq \alpha, \quad (6b)$$

where  $\bar{E}_{22i}$  is the initial transverse modulus,  $E^*$  the uncured transverse modulus,  $\alpha^*$  the degree of cure at initial

transverse modulus development, and  $a_0$ ,  $a_1$  and  $a_2$  are the transverse modulus modelling parameters.

The longitudinal modulus of elasticity  $\bar{E}_{11}$  and major Poisson's ratio  $\bar{\nu}_{12}$  can be expressed as linear functions of  $\alpha$ , viz.

$$\bar{E}_{11}(\alpha) = \bar{E}_{11i} + (\bar{E}_{11f} - \bar{E}_{11i})\alpha, \quad (7)$$

$$\bar{\nu}_{12}(\alpha) = \bar{\nu}_{12i} + (\bar{\nu}_{12f} - \bar{\nu}_{12i})\alpha, \quad (8)$$

where  $\bar{E}_{11i}$  is the uncured longitudinal modulus,  $E_{11f}$  the fully cured longitudinal modulus,  $\bar{\nu}_{12i}$  is the uncured major Poisson's ratio and  $\bar{\nu}_{12f}$  is the fully cured major Poisson's ratio.

The elastic compliances are given by

$$\bar{S}_{22i}(\alpha) = \frac{1}{\bar{E}_{22i}(\alpha)}, \quad (9)$$

$$\bar{S}_{11}(\alpha) = \frac{1}{\bar{E}_{11}(\alpha)}, \quad (10)$$

$$\bar{S}_{12}(\alpha) = \frac{-\bar{\nu}_{12}(\alpha)}{\bar{E}_{11}(\alpha)}, \quad (11)$$

where  $\bar{S}_{11}$  is the compliance, and  $\bar{S}_{12}$  is the longitudinal shear compliance. The in-plane stiffness constants  $\bar{Q}_{ij}$  are related to compliances by the following expressions:

$$\bar{Q}_{11}(\alpha, t) = \frac{\bar{S}_{22}(\alpha, t)}{\bar{S}_{11}(\alpha)\bar{S}_{22}(\alpha, t) - \bar{S}_{12}^2(\alpha)}, \quad (12)$$

$$\bar{Q}_{22}(\alpha, t) = \frac{\bar{S}_{11}(\alpha)}{\bar{S}_{11}(\alpha)\bar{S}_{22}(\alpha, t) - \bar{S}_{12}^2(\alpha)}, \quad (13)$$

$$\bar{Q}_{12}(\alpha, t) = \frac{-\bar{S}_{12}(\alpha)}{\bar{S}_{11}(\alpha)\bar{S}_{22}(\alpha, t) - \bar{S}_{12}^2(\alpha)}. \quad (14)$$

### 3. Residual stresses and moments

A residual stress–strain analysis for unsymmetric cross-ply laminates are given next based on computing chemical and thermal strains in the first step. The longitudinal chemical strain can be taken to be zero since the fibres do not experience chemical strains during the process cycle. Experimental results obtained by White and Hahn [15] indicate that an accurate representation of the transverse chemical strains is given by expressions of the form

$$e_2^c = \beta_1 + \beta_2 10^{(\beta_3 \alpha)}, \quad \alpha \leq \alpha^c, \quad (15a)$$

$$e_2^c = e_2^{cf}, \quad \alpha > \alpha^c, \quad (15b)$$

where  $e_2^c$  the laminate chemical strains,  $\beta_1$ ,  $\beta_2$ , and  $\beta_3$  are empirical transverse chemical strain coefficients and exponent, respectively,  $\alpha^c$  the degree of cure when chemical

shrinkage is complete and  $e_2^{cf}$  is the final transverse chemical shrinkage strain.

The thermal strains  $e_i^T$  can be obtained from

$$e_i^T = \alpha_i(T - T_0), \quad (16)$$

where  $\alpha_i$  is the thermal expansion coefficient and  $T_0$  is the initial stress free temperature. Experimental results of White and Hahn [15] shows that  $\alpha_i$ 's do not depend on the degree of cure.

Once the strains are known, the residual stresses can be computed from

$$\sigma_i = \bar{Q}_{ij}(\epsilon_j - e_j), \quad (17)$$

where  $\sigma_i$ 's are the laminate stresses and  $\epsilon_j$ 's are the mechanical strains. In Eq. (17) the strains  $e_j$  are given by

$$e_j = e_j^c + e_j^T \quad (18)$$

where  $e_j^c$ 's are the chemical strains, and  $e_j^T$ 's the thermal strains. The total strains  $\epsilon_j^0$  at a distance  $z$  from the mid-plane can be obtained from the mid-plane strains  $\epsilon_j^0$  and curvatures  $k_j$  using the relation

$$\epsilon_j = \epsilon_j^0 + zk_j. \quad (19)$$

For a cross-ply laminate with a stacking sequence  $[0_n/90_n]$ , the longitudinal and transverse stresses in the individual plies are given by (see Ref. [16])

$$\begin{aligned} \sigma_1^0 &= \bar{Q}_{11}\epsilon_1 + \bar{Q}_{12}\epsilon_2 - (\bar{Q}_{11}e_1 + \bar{Q}_{12}e_2), \\ \sigma_2^0 &= \bar{Q}_{12}\epsilon_1 + \bar{Q}_{22}\epsilon_2 - (\bar{Q}_{12}e_1 + \bar{Q}_{22}e_2), \\ \sigma_1^{90} &= \bar{Q}_{22}\epsilon_1 + \bar{Q}_{12}\epsilon_2 - (\bar{Q}_{22}e_2 + \bar{Q}_{12}e_1), \\ \sigma_2^{90} &= \bar{Q}_{12}\epsilon_1 + \bar{Q}_{11}\epsilon_2 - (\bar{Q}_{12}e_2 + \bar{Q}_{11}e_2), \end{aligned} \quad (20)$$

where the superscripts 0 and 90 refer to the respective plies, and the subscripts to the 1 (longitudinal) and 2 (transverse) directions.

During the curing process of composite materials, the laminate is usually kept flat until the completion of the cure cycle. By using the laminated plate theory subject to the constraints that the stress resultants  $N_i = 0$  and the curvatures  $k_i = 0$ , the mid-plane strains can be computed as

$$\epsilon_j^0 = \frac{(\bar{Q}_{11} + \bar{Q}_{12})e_1 + (\bar{Q}_{12} + \bar{Q}_{22})e_2}{\bar{Q}_{11} + \bar{Q}_{22} + 2\bar{Q}_{12}}. \quad (21)$$

Similarly the moment resultants are given by (see Ref. [11])

$$M_1 = -M_2 = \frac{h^2(\bar{Q}_{12}^2 + \bar{Q}_{11}\bar{Q}_{22})(e_1 - e_2)}{\bar{Q}_{11} + \bar{Q}_{22} + 2\bar{Q}_{12}}, \quad (22)$$

where  $M_i$ 's are the bending moments and  $h$  is the laminate half thickness.

Polymer materials are known to exhibit time-dependent mechanical behaviour especially at high temperature. For linear thermorheologically simple materials a single temperature dependent shift factor,  $a_T(T)$ , can be used to predict the transient thermal response [17]. For

viscoelastic materials, mechanical response is history dependent and involves the use of reduced times,  $\xi(t)$  and  $\xi(\tau)$ . These reduced times can be found from the shift factors as

$$\begin{aligned}\xi(t) &= \int_0^t \frac{ds}{a_T[T(s)]}, \\ \xi(\tau) &= \int_0^\tau \frac{ds}{a_T[T(s)]}.\end{aligned}\quad (23)$$

The shift factor is modeled considering a temperature dependence only according to the general equation (see Ref. [11])

$$a_T(T) = \exp\left(\frac{B_1}{T} - B_2\right), \quad (24)$$

where  $B_1$  and  $B_2$  are empirical shift factor modeling parameters.

Furthermore, the quasi-elastic method as developed by Schapery [18] is used in the development of the viscoelastic residual stress model. When the quasi-elastic method is used, the viscoelastic resultant moment can be approximated by substituting the time-dependent stiffnesses for elastic stiffnesses in Eq. (22) and making use of the convolution integral. The results are given by

$$\begin{aligned}M_1(t) &= -M_2(t) \\ &= h^2 \int_0^t F\{\alpha(\tau), [\xi(t) - \xi(\tau)]\}(\dot{e}_1(\tau) - \dot{e}_2(\tau))d\tau,\end{aligned}\quad (25)$$

where a dot denotes differentiation and  $F$  is given by

$$F(\alpha, t) = \frac{[\bar{Q}_{12}^2(\alpha, t) + \bar{Q}_{11}(\alpha, t)\bar{Q}_{22}(\alpha, t)]}{[\bar{Q}_{11}(\alpha, t) + \bar{Q}_{22}(\alpha, t) + 2\bar{Q}_{12}(\alpha, t)]}. \quad (26)$$

Since Eq. (25) is a history dependent integral, the degree of cure,  $\alpha(\tau)$ , is represented as a function of  $\tau$  and not the current time,  $t$ . Eq. (25) is solved by discretising the time domain into  $n$  equal portions  $\Delta t$  with the initial time  $t_i = 0$  and the current time  $t_n = t$ :

$$\begin{aligned}M_1(t) &\approx \frac{1}{2}h^2 \sum_{k=1}^n \{F[\alpha(t_{k+1}), \xi(t_{n+1}) - \xi(t_{k+1})] \\ &\quad + F[\alpha(t_{k+1}), \xi(t_{n+1}) - \xi(t_k)]\}[\Delta e(t_{k+1}) - \Delta e(t_k)],\end{aligned}\quad (27)$$

$F$  is calculated using the time-dependent stiffnesses obtained from Eqs. (12)–(14) while  $\Delta e$  is calculated from Eq. (23) (Strain history modeling), thus facilitating the calculation of the residual moments.

#### 4. Numerical results and discussion

The model input parameters and values based on the experimental study conducted by White and Hahn [15] are given below.

Length of cure	Variable, dependent on temperature gradients
First dwell temperature	131°C
Second dwell temperature	181°C
Cure kinetics constant, $K1$	$(1.98) \exp(-2770/T) \text{ s}^{-1}$
Cure kinetics constant, $m1$	$1.17 - (1.74e-3)(T)$
Cure kinetics constant, $n1$	$199 - (0.415)(T)$
Cure kinetics constant, $K2$	$6550 \exp(-7040/T) \text{ s}^{-1}$
Cure kinetics constant, $m2$	0
Cure kinetics constant, $n2$	$13.2 - (0.025)(T)$
Cure kinetics constant, $K3$	$81.9 \exp(-5340/T) \text{ s}^{-1}$
Cure kinetics constant, $m3$	0
Cure kinetics constant, $n3$	$131 - (0.558)(T) + (6e-4)(T^2)$
First break point as a function of dwell temperature $Tc2$	$(3.44e-12)(10^{(0.022)(Tc2)})$
Second break point	$-25.7 + (0.11)(Tc2) - (1.15e-4)(Tc2)^2 - 0.3e-6 \text{ m}^\circ\text{C}$
Longitudinal thermal expansion coefficient, $\alpha_1$	
Transverse thermal expansion coefficient, $\alpha_2$	$30e-6 \text{ m}^\circ\text{C}$
Initial stress-free temperature, $T_0$	290 K
Chemical strain coefficients, $\beta_1, \beta_2, \beta_3$	0.005, $-525e-5$ , 1 respectively
Degree of cure when chemical shrinkage is complete, $\alpha^c$	0.81
Final transverse chemical shrinkage strain, $e_2^{c,f}$	-0.029
Uncured and fully cured transverse creep coefficients, $D_i, D_f$	$2.72e-9$ , $-2.72e-9$ , respectively
Uncured and fully cured transverse creep exponents, $q_i, q_f$	0.123, 0.24, respectively
Shift factor constants, $B_1, B_2$	6190, 20.3, respectively
Initial transverse modulus modelling coefficients, $a_0, a_1, a_2$	-214 GPa, 451 GPa, -228 GPa, respectively
Uncured transverse modulus, $E^*$	2 GPa
Cure at initial transverse modulus development, $\alpha^*$	0.82
Uncured and fully cured longitudinal modulus, $E_{11i}, E_{11f}$	114 GPa, 183 GPa, respectively

Uncured and fully cured major Poisson's ratio, $v_{12i}$ , $v_{12f}$	0.4, 0.31, respectively
Half thickness of laminate, $h$	0.765 mm
Minor Poisson's ratio, $v_{21}$	0.35
Final shear modulus, $G$	15 GPa

#### 4.1. Residual stress vs. temperature gradient

When considering the input requirements of the chosen process model, it is observed that the only parameters that the engineer has control over are the initial stress free temperature and the applied temperature profile. The rest of the inputs involve empirically determined material constants. Next the effect of these controllable variables on the residual stress is studied.

The results of White and Hahn [7,15] indicate that the residual moment and hence residual stress profiles follow closely the applied temperature profile. Further, the greatest increase in residual moment occurs during the cooldown phase of the cure cycle. Thus the temperature gradient of cooldown from the second dwell temperature to room temperature is the major contributing factor to the resulting residual stresses. Thus a decrease in residual stress could be expected with increasing gradient at the cooldown phase (i.e., decreasing negative gradient).

However, the validity and the range of applicability of this observation should be checked. For this purpose, the residual stress on the upper surface of the composite is plotted against the cooldown gradient in Fig. 1. It is observed that the residual stress decreases initially with increasing gradient up to a point of optimum gradient where the stress is the lowest and thereafter begins increasing with further increase in the gradient. The optimum gradient for the cooldown phase in this case is found to be  $-0.0523$ . The change in stress observed is substantial as it decreases from a maximum of 183 MPa

to a minimum of 68 MPa, i.e., a 63% decrease. Such a dramatic decrease in residual stress will definitely increase the ultimate strength of the composite quite significantly.

It is clear that the determination of this optimum gradient point is essential in order to reduce the residual stresses that develop during curing process.

Next the effect of the first to second dwell temperature rise on the residual stress is investigated. The results of White and Hahn [7,15] indicate that there is also an increase in residual moment (stress) in the period of temperature rise from the first to second dwell temperatures. However, it is necessary to determine the dependence of the residual stresses on the gradient of the temperature rise in this phase to establish the optimum temperature profile. The results of this investigation are shown in Fig. 2 which indicates that the residual stress decreases with increasing gradient. The lowest stress achieved is 63 MPa at a gradient that constitutes an almost instantaneous jump from the first dwell temperature to the second dwell temperature.

The minimum gradient investigated, 0.0104, produces a stress of 89.5 MPa. This results in a possible decrease of 30% in the residual stress from the maximum value investigated. Such a decrease is quite significant considering that the temperature gradient that yields it is relatively small.

The results obtained so far has a double advantage; the steep gradient firstly reduces the residual stresses and secondly shortens the processing time of the cure cycle. The latter advantage obviously impacts on the production rate of the manufacturing concern and hence on the manufacturing cost of the product.

Next the effect of the rate of increase of temperature to the first dwell on the residual stress is considered. During this phase the stress remains fairly flat. However, the fact that the residual stress response is history dependent (viscoelastic) indicates that every change in the applied temperature can have an effect on the final

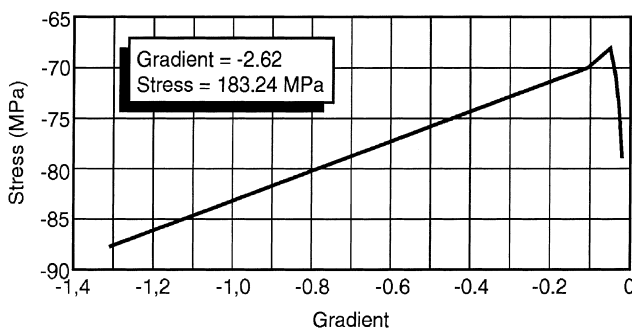


Fig. 1. Graph of residual upper surface stress vs. gradient of cooldown.

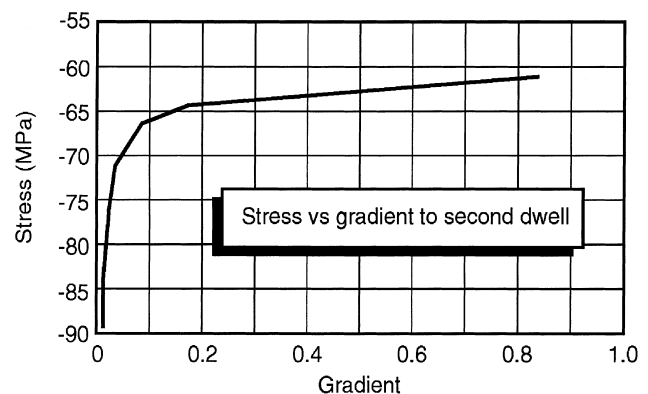


Fig. 2. Residual upper surface stress vs. gradient of temperature rise between first and second dwell temperatures.

residual stress value, even if there might be no effect at the time of change.

This point is investigated in Fig. 3 by plotting the final value of the residual stress against the temperature gradient to the first dwell temperature. Fig. 3 is similar to Fig. 2 in that the residual stress decreases with increasing gradient. Thus, the best gradient for the initial temperature rise would be an instantaneous jump from room temperature to the first dwell temperature. However, as this would not be a practical solution to implement, the recommended one is a gradient which is as high as practically possible. The minimum stress achieved is 57 MPa at a gradient of 1.783 and the maximum stress is 91 MPa resulting from a gradient of 0.0209. This amounts to a possible decrease in residual stress of 37% from the maximum value investigated. The result observed in Fig. 3 also has the second advantage of reducing the processing time as well as the residual stresses.

#### 4.2. Pre-heating response

Normally, in most industrial cure cycles, the initial stress free temperature is the room temperature. However, the materials can be pre-heated separately before being combined to form the composite and allowing curing to commence. This is possible because curing only takes place significantly above a specific temperature. Temperatures below this critical temperature do not facilitate the cross-linking reaction associated with curing and hence the cure is not substantial. Investigation of the effect of pre-heating for the present problem, however, yielded only a 1.3% decrease in stress for a pre-heating temperature difference of 104°C. The additional expenses incurred in pre-heating the individual materials cannot be justified by the small decrease in residual stresses. Hence, pre-heating the

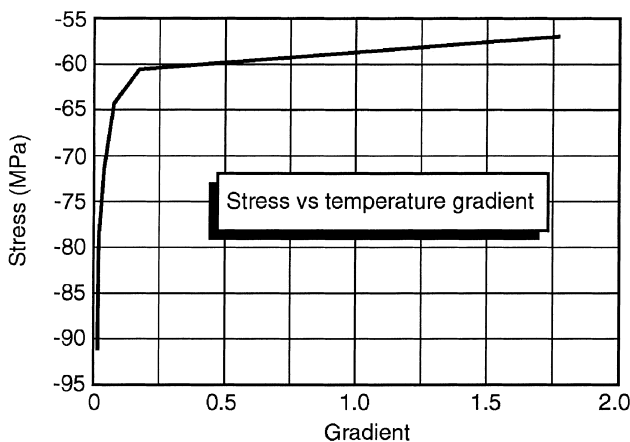


Fig. 3. Residual upper surface stress vs. gradient of temperature rise to first dwell temperature.

composite component materials is not considered to be a practical option to decrease the processing induced residual stresses.

It is noted that the trends observed in Sections 4.1 and 4.2 are evident in both plies of the laminate. The figures referred to, however, apply to the upper surface of the 0° orientated ply.

#### 4.3. Optimising the cure cycle

Having obtained the required tools, it now becomes possible to develop an optimised curing cycle. Combining the results of Figs. 1–3 and disregarding any pre-heating of component materials yields Fig. 4(a). The increase in temperature from the initial stress free temperature to the first dwell temperature is virtually instantaneous and so is the transition from the first dwell temperature to the second dwell temperature. The rate of temperature decrease from the second dwell back to room temperature is at the optimum gradient of -0.0523 found in Fig. 1.

An important process parameter is obviously the degree of cure which directly affects the quality of the final product by virtue of determining the stiffness and strength constants. This aspect is investigated in Fig.

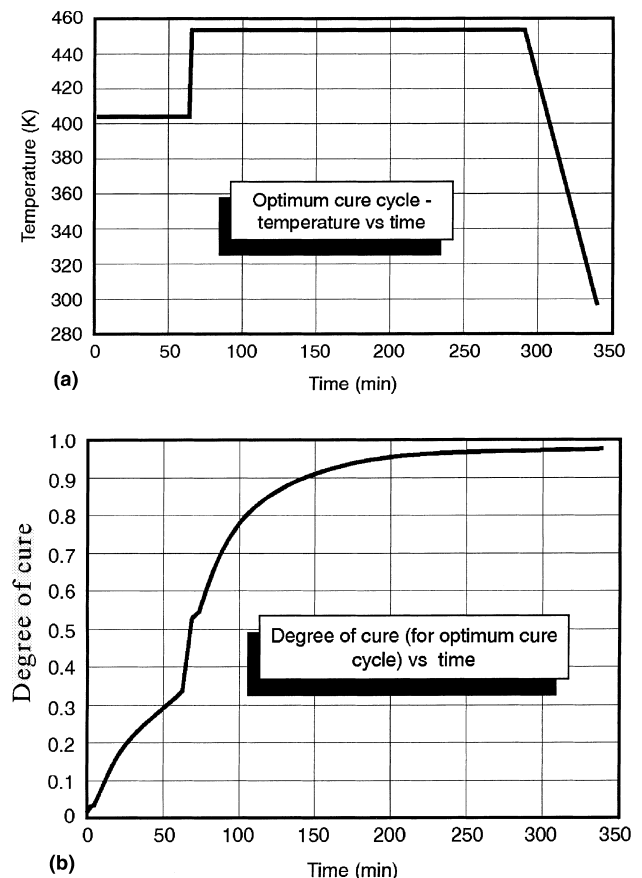


Fig. 4. (a) Optimum cure cycle and (b) degree of cure.

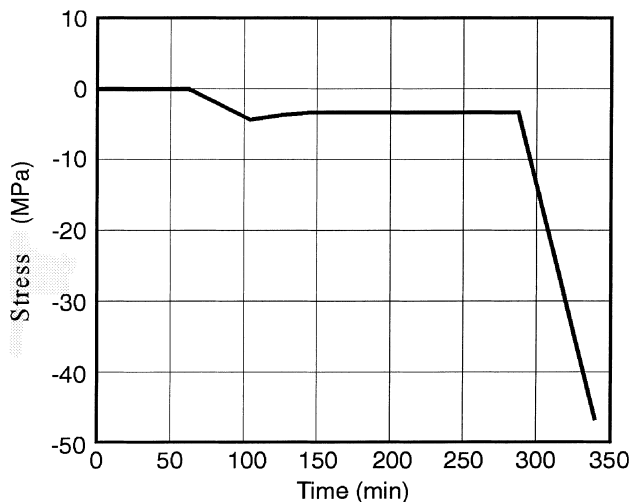


Fig. 5. Residual upper surface stress vs. time with the optimum cure cycle.

4(b) by plotting the degree of cure against time for the temperature profile shown in Fig. 4(a). It is observed that the degree of cure is higher than 99% which is the recommended minimum value in practice.

The resulting stress history is shown in Fig. 5. When this is compared to the resultant stresses resulting from the manufacturer's recommended cure cycle (MRC) the significance in the decrease becomes evident. The MRC implemented cure cycle is simulated using data from White and Hahn [15] and yields a final residual stress of 71.2 MPa while the optimised cure cycle results in a final stress of 48 MPa, i.e., a 32.5% decrease. The other important reduction is in the processing time which decreases from 425 min for the implemented cycle to 339 min for the optimised cure cycle. This difference of almost a hour and a half and corresponds to a 20% increase in productivity.

The optimised cycle may, however, be difficult to implement due to the almost instantaneous temperature changes as mentioned before. In these cases, the most rapid practical temperature change should be applied.

## 5. Conclusion

A numerical simulation study was given for polymer matrix composites with the aim of studying and minimizing thermally-induced residual stresses during the curing process using a viscoelastic model. The analysis conducted enabled a better understanding of the behaviour of the resulting residual stresses to changes in the cure cycle. This understanding was used to observe that the temperature gradients of first and second dwell times and the cooldown period have strong influence on the residual stresses. It was concluded that by choosing

these gradients in an optimum manner, the residual stresses can be reduced substantially.

The optimum temperature profile determined by the above procedure yielded not only reduced residual stresses, but also shorter curing times for fully cured components. Optimisation of the cure cycle has brought an important realisation: Possibly most industrial cycles are not achieving the lowest possible residual stresses. Hence the materials produced are not able to perform at their maximum obtainable strength. Implementation of an optimised cycle can result in higher strength materials by virtue of reduced residual stresses as well as reduced composites manufacturing costs due to higher temperature gradients. In the present example, the residual stress was reduced by more than 30% and the cure cycle by about 20%. Both these reductions are significant from a practical point of view for the curing processes of composite materials.

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